BULLETIN

OF THE INSTITUTE OF METALS

VOLUME 3 JULY 1956 PART 11

INSTITUTE NEWS

Institute's New Headquarters

Second List of Donations to the House Fund

The Council gratefully acknowledges the following additional donations to the House Fund:

	£	s.	d.
*High Duty Alloys, Ltd.	3500	0	0
*Enfield Rolling Mills, Ltd.	700	0	0
Davy and United Engineering Co., Ltd.	250	0	0
Goodlass Wall and Lead Industries, Ltd.	250	0	0
Glynwed Tubes, Ltd.	105	0	0
Loewy Engineering Co., Ltd., The	100	0	0
Pyrotenax, Ltd.	100	0	0
Nickel Silver Association	52	10	0
Electric Furnace Co., Ltd.	30	0	0
*Hume-Rothery, Dr. W.	12	15	6
Howard, Mr. J. C.	10	0	0
Robertson, Dr. J. M.	8	8	0
Finniston, Dr. H. M.	5	5	0
Pumphrey, Dr. W. I.	4	4	0
Doyle, Dr. W. M.	2	2	0
de Winton, Mr. C. B. H.	2	0	0
Clotworthy, Mr. S. E.	1	1	0
Fletcher, Mr. L.	1	1	0
Thornely, Mr. B. N. H.	1	1	0
Tucker, Mr. G. E. G.	1	0	0

* Donation under covenant. The amount stated is the total to be received, including tax recoverable at the present standard rate, where applicable.

Autumn Meeting, Germany, 17–25 September

All members resident in Europe should now have received a copy of the final programme of the Autumn Meeting to be held in Germany from Monday 17 to Tuesday 25 September 1956. Other members may obtain a copy on application to the Secretary. The details do not differ greatly from those given in the provisional programme published in the March issue of the *Bulletin* (pp. 52–53).

Second World Metallurgical Congress

With the June issue of the *Journal*, members should have received details of the Second World Metallurgical Congress, to be held in Chicago, Ill., on 2–8 November 1957. Any members who are intending to take part in the Congress are reminded that they should complete without delay the preliminary application form.

Honorary Life Members

By a special resolution, the Council has conferred Honorary Life Membership on all Original Members who were still members on 25 April 1956 in recognition of their continued support of the Institute since its foundation. The Original Members concerned are:

ANDREW, Professor Emeritus J. H.
BAMFORD, C. C.
CLAMER, Dr. G. H.
COOKSON, Clive.
DESCH, Dr. C. H. (*Fellow*).
DRURY, H. J. J.
FERON, Albert.
GARNHAM, F. M.
GREENWOOD, H. W.
GWYER, Dr. A. G. C.
HIGHTON, D. C.
HOLT, H.
HUMPHREYS, T. C.
HUTTON, Professor Emeritus R. S. (*Fellow*).
JOHNSON, A. L.
KAMPS, F. J. H.
KEMP, J. F.
MORCOM, E. L.
PRIMROSE, H. S.
RADLEY, W. A.
ROBERTSON, W. H. A. (*Fellow*).
SCOTT, G. Shaw (*Secretary Emeritus*) (*Fellow*).
SELIGMAN, Dr. Richard (*Fellow*).
SPITTLE, A.
WEBB, A. J.
WEIR OF EASTWOOD, The Right Hon. the Viscount.

Election of Members

The following 2 Overseas Sustaining Members, 10 Ordinary Members, 2 Junior Members, and 11 Student Members were elected on 29 May 1956:

As Overseas Sustaining Members

CONSOLIDATED MINING AND SMELTING COMPANY OF CANADA, LTD., Trail, B.C., Canada.

CRANE, LTD., 1170 Beaver Hall Square, Montreal, P.Q., Canada.

As Ordinary Members

EDWARDS, Gordon W., Vice-President in Charge of Sales, Kensico Tube Company, Inc., Hubbell's Crossroad, Mount Kisco, N.Y., U.S.A.

GAUNT, John Richard, B.Sc., A.M.I.Mech.E., Managing Director, J. R. Gaunt and Son, Ltd., Warstone Parade Works, Birmingham 18.

HUG, Hans, Chief Metallurgist, Forschungsinstitut, Aluminium Industrie A.G., Neuhausen-am-Rheinfall, Switzerland.

MARSHALL, Peter Robert, B.Sc., Ph.D., F.R.I.C., General Manager, Metal and Plastic Compacts, Ltd., Montgomery Street, Birmingham 11.

MULLER, Paul, Doct.ès Sc., Vize-Direktor, Aluminium Industrie A.G., Avenue d'Ouchy 61, Lausanne, Switzerland.

NAST, Thomas, A.B., President, Kensico Tube Company, Inc., Hubbell's Crossroad, Mount Kisco, N.Y., U.S.A.

OSWALD, John McLay, B.Sc., A.R.T.C., A.I.M., Chief Metallurgist, Pressed Steel Co., Ltd., Linwood, near Paisley, Renfrewshire.

PECK, William Charles, M.Sc., M.I.Chem.E., F.R.I.C., A.M.I.Mech.E., Managing Director, Apex Construction, Ltd., 15 Soho Square, London, W.1.

TAYLOR, Lancelot Frederick, G.I.Mech.E., Assistant Engineer, Copper and Alloys, Ltd., Greets Green Road, West Bromwich, Staffs.

TYRWHITT-DRAKE, Thomas William, M.A., Personal Assistant to the Technical Director, Henry Wiggin and Co., Ltd., Wiggin Street, Birmingham 16.

As Junior Members

HOLMES, Fred, B.Sc., Assistant Metallurgist, Northern Aluminium Company, Ltd., Middlemore Road, Handsworth, Birmingham 21.

ROSE, Joseph Oswald, Metallurgical Laboratory Assistant, Mullard Magnetic Components, Crossens, Southport, Lancs.

As Student Members

ASTON, Malcolm Arthur James, Metallurgist, Bayliss, Jones and Bayliss, Ltd., Cable Street, Wolverhampton.

CONYNGHAM, Maxwell George, Junior Metallurgist, 8 Malleny Street, Ashbury, Sydney, N.S.W., Australia.

DAVIS, Peter John, Trainee Metallurgist, 62 Douglas Street, Blacktown, N.S.W., Australia.

DERRICOTT, Robert Thomas, Student, Department of Metallurgy, College of Technology, Birmingham.

HOBBS, Dennis, Undergraduate, Department of Metallurgy, University College, Swansea.

KIMURA, Hiroshi, B.S., Graduate Student, School of Metallurgical Engineering, University of Pennsylvania, Philadelphia, Pa., U.S.A.

PEARCE, John Hardy, Assistant Experimental Officer, Metallurgy Division, Atomic Energy Research Establishment, Harwell, Berks.

DE PLATER, Roy C. M., Trainee Metallurgist, 141 Burnett Street, Parramatta, N.S.W., Australia.

PITT, Terence John, Student, Department of Metallurgy, College of Technology, Birmingham; Student Apprentice, G. K. N. Group Research Laboratories, Wolverhampton.

SARTELL, Jack Albert, B.Met.E., M.S., Graduate Student, Department of Mining and Metallurgy, University of Wisconsin, 1509 University Avenue, Madison, Wis., U.S.A.

WIESE, John Rudolf, B.S., Graduate Student, School of Metallurgical Engineering, University of Pennsylvania, Philadelphia, Pa., U.S.A.

PERSONAL NOTES

MR. W. E. BALLARD has been elected a Vice-President of the Institution of Metallurgists.

DR. F. P. BOWDEN, Reader in Physical Chemistry in the University of Cambridge, received the C.B.E. in the Birthday Honours.

MR. P. J. BOWLES has been awarded the Ph.D. degree of London University.

DR. H. W. CLARKE, Chairman of the Aluminium Industry Council, received a Knighthood in the Birthday Honours.

MR. J. CROWTHER, Chief Metallurgist to James Booth and Co., Ltd., received the M.B.E. in the Birthday Honours.

DR. R. W. DIAMOND is to be the President of the Sixth Commonwealth Mining and Metallurgical Congress (see May *Bulletin*, p. 81). He recently retired from the position of Executive Vice-President, Western Region, of The Consolidated Mining and Smelting Company of Canada, Ltd.

DR. H. K. HARDY is now Deputy Head of Springfields Laboratories, U.K. Atomic Energy Authority (Industrial Group), Springfields, Preston, Lancs.

PROFESSOR A. J. MURPHY, Principal of the College of Aeronautics, has been elected a Vice-President of the Institution of Metallurgists.

MR. R. L. PRAIN, Chairman and President of the Rhodesian Selection Trust, Ltd., and Roan Antelope Copper Mines, Ltd., received a Knighthood in the Birthday Honours.

DR. C. SYKES, Managing Director of Thos. Firth and John Brown, Ltd., received the C.B.E. in the Birthday Honours.

MR. H. H. SYMONDS has been elected President of the Birmingham Metallurgical Society for 1956–57.

DR. H. G. F. WILSDORF has left the National Physical Laboratory, Pretoria, and joined the staff of the Solid-State Physics Department of the Franklin Institute Laboratories for Research and Development, Philadelphia, Pa.

DR. P. A. YOUNG has left the Imperial Smelting Corporation, Ltd., Avonmouth, to become Manager of Research and Development to the Head, Wrightson Group.

Death

The Editor regrets to announce the death of:

DR. HENRY LEPP at the Château de Siorac, France, on 25 May 1956, after a long illness.

NEWS OF KINDRED SOCIETIES

Société Française de Métallurgie

The Autumn Meeting of the Société Française de Métallurgie will be held in Paris from 22 to 27 October 1956. An advance programme will be available in July, and members of the Institute of Metals will be warmly welcomed at the meeting. The address of the Société is 25, rue de Clichy, Paris (9e).

Sociedad Argentina de Metales

A metallurgical society was established in Argentina at the end of last year, with the title "Sociedad Argentina de Metales". The first president is Ingeniero Antonio Sturla, and the secretary Professor Jorge A. Sábato, who is Corresponding Member to the Council of the Institute of Metals for Argentina. The offices of the Sociedad are at Diag. Pte. Roque Saenz Peña 555, 8° Piso, Buenos Aires.

Student's Essay

A CONSIDERATION OF THE CHANGES AND MECHANISMS INVOLVED DURING THE ANNEALING OF COLD-WORKED METALS *

B13

By M. DEIGHTON,† B.Sc.

SYNOPSIS : The various changes occurring on annealing, i.e. recovery, polygonization, recrystallization in situ, *and primary and secondary recrystallization, are defined and discussed in turn. It is suggested that recovery is due to interaction of dislocations of opposite sense, that polygonization is a stress-induced process occurring during deformation, and that recrystallization* in situ *is a growth of these subgrains. Primary recrystallization is a preferential growth of a few subgrains and thus is not easily distinguished from secondary recrystallization. Theories of nucleation are considered, and it is concluded that no nucleus as such is formed ; rather, certain subgrains grow preferentially after an initial period of strain-relief by recovery.*

Introduction

WHEN a metal is cold worked, its hardness is usually increased and the metal is said to become " work-hardened ". If the metal be then annealed, the strain-hardening is removed and

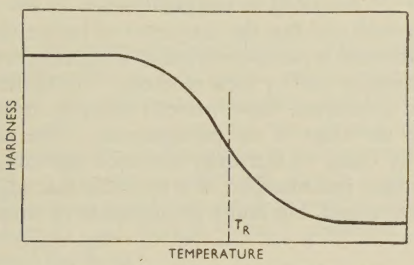

FIG. 1.—Variation of Hardness with Annealing Temperature.

the metal attains its original hardness. This phenomenon has been utilized in industrial processes for many years and is generally known as " recrystallization ", a process of replacing deformed grains by soft, strain-free ones.

One of the earliest investigations of the subject was made by Carpenter and Elam,[1] who enunciated what may be termed the general laws of recrystallization, which express the relationship between grain-size, deformation, and temperature. These have recently been reviewed and verified by Williams and Eborall.[2]

Some of the literature suggests that the changes occur at a temperature known as the " recrystallization temperature ", which can be derived from a diagram of the type shown in Fig. 1, but a closer study of the variables, i.e. deformation and time, shows that T_R, the recrystallization temperature, is a function of annealing time and amount of cold work. Thus, the concept of a " recrystallization temperature " is purely arbitrary and has no fundamental significance.

During annealing, a deformed metal undergoes changes and may pass through several stages before becoming fully softened. These stages are variously known as " recovery ", " polygonization ", " recrystallization *in situ* ", " primary recrystallization ", and " secondary recrystallization ". They are defined as follows :

" Recovery " is the recovery of physical properties without observable structural changes.

" Polygonization " is the formation of slightly misaligned blocks in the existing grains.

" Recrystallization *in situ* " is the growth of these blocks.

" Primary recrystallization " is the disappearance of deformed material and appearance of new grains.

" Secondary recrystallization " is the migration of existing interfaces.

During cold working, up to 10% of the energy expended in causing the deformation is stored in the metal and provides the driving force for the changes observed on annealing. Taylor and Quinney[3,4] found that heat was evolved during annealing in the stages ascribed to recovery and recrystallization.

Recovery

When zinc single crystals are deformed simply on one glide system, the work-hardening introduced is rapidly removed at room temperature. No interfaces are introduced into the crystals during deformation or annealing. The recovery is homogeneous on a microscopic scale and involves no interface migration. If, however, the deformation is complex and involves bending, only part of the work-hardening is recoverable. For this reason recovery is never complete in polycrystals.

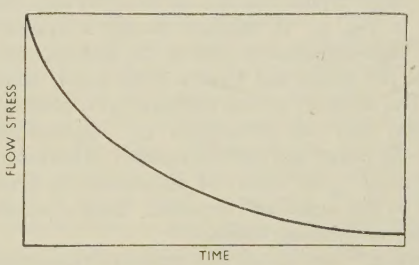

FIG. 2.—Variation of Flow Stress with Annealing Time.

Fig. 2 shows a typical recovery curve, where the initial rate is greatest and decreases with time. This is contrary to a mechanism involving nucleation and growth, where the initial rate is least. The recovery process is dissimilar to recrystallization and, thus, can not be due to interface migration.

The detection of recovery is not restricted to mechanical properties, but can be followed by changes in other structure-sensitive properties. Changes in electrical resistivity have been

* Awarded the Students' Essay Prize of the Institute of Metals for 1955. † King's College, University of Durham, Newcastle-upon-Tyne.

employed by Tammann and Dreyer [5] as a means of following the recovery of zinc and magnesium. Recrystallization appeared to be preceded by recovery in an isochronal anneal, but it may be that recovery is the more rapid process at lower temperatures and that softening occurs preferentially by this process.

The amount of recovery of work-hardening and other properties increases with decreasing deformation temperatures. The isothermal rate of recovery of each property change is greatest initially and gradually decreases. Over some of the range, the recovery of flow stress and stored energy is a linear function of the logarithm of annealing time, so that, here, the rate of recovery is inversely proportional to annealing time.

Polygonization and Recrystallization *in situ*

When single crystals are bent, the Laue asterisms obtained from a diffraction pattern become spotty on annealing, and the metal is said to have recrystallized *in situ*. The break-up of the asterisms may be attributed to the atomic planes becoming short, straight portions delineating the original curve. Andrade [6, 7] noticed these effects when sodium and potassium extended at room temperature showed spotted asterisms, but

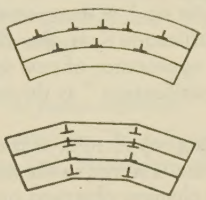

FIG. 3.—Schematic Representation of Bent and Bent and Annealed Crystal.

not at lower temperatures. Zinc single crystals and polycrystals have been observed to recrystallize *in situ* before nucleation.[8]

Cahn [9] has studied recrystallization *in situ* and established that bent single crystals are the most susceptible. The microstructures of bent and annealed single crystals reveal many small crystallites separated by straight boundaries, perpendicular to the slip planes. The small blocks are only slightly disoriented, and it may be supposed that the boundaries are arrays of edge dislocations.

A bent single crystal contains an excess of edge dislocations, as shown in Fig. 3. If sufficient energy is available, as in a relatively high-temperature anneal, the dislocations can climb from their slip planes and form a stable array, as indicated in Fig. 3. The accuracy of this explanation is confirmed by the observations that the boundaries are perpendicular to the operative slip planes and that the number of boundaries, which is proportional to the excess of dislocations, is inversely proportional to the radius of curvature. Such a process leads to greater thermodynamic stability.[10]

Heidenreich [11] has shown that grains of aluminium, heavily cold worked by hammering at room temperature, consist of slightly disoriented, imperfect subgrains having an average diameter of about 2 μ. Heidenreich concludes that these subgrains, immediately after deformation, possess considerable strain energy. However, on standing at room temperature, this strain energy disappears, and the subgrains become more perfect during recovery. The subgrain size remains constant throughout recovery, even at elevated temperatures. This type of substructure has been observed in several metals.[12–15]

According to Beck,[16] subgrains are formed in metallic crystals on plastic deformation, presumably by bending, even at temperatures as low as one-tenth of the absolute melting point. The only deformation so far known not to be connected with subgrain formation is pure shear, without bending, in single crystals. Accepting the dislocation model of sub-boundaries, it seems that the arrangement must be stress-induced, since thermal movement of dislocations, at temperatures as low as one-tenth of the absolute melting point, would be too slow to account for the changes in structure observed so shortly after deformation. The subgrains are produced by the deformation itself and not from the deformed state by recovery.

Annealing, under suitable conditions, may produce subgrain growth, leading to a decrease in total sub-boundary energy and increased thermodynamic stability. Subgrain growth is often prevented, particularly in deformed polycrystalline metals, by premature onset of recrystallization, which wipes out the substructure. The rate of subgrain growth is greatly increased by an external stress applied during annealing, as in creep testing, where subgrain structures of considerable size are found. Large work-hardening effects are connected with the substructure produced in plastic deformation, and subgrain growth during annealing may account for at least part of the strain release. The extent of relief by this mechanism depends mainly on its rate in relation to that of recrystallization.

In view of the above evidence of a polygonized structure arising from deformation, it seems that what Cahn [9] and others [6–8, 17, 18] described as recrystallization *in situ* is merely subgrain growth and that the time interval before observation of recrystallization *in situ* is occupied by subgrain growth to the limit of resolution used by these workers. It is significant that, using finer techniques, these workers disagree among themselves as to the onset of the phenomenon. The mechanism suggested by Cahn of thermally activated diffusion of dislocations becomes inadmissible; it is probable that edge dislocations are rearranged, but under the influence of stress.

Recrystallization

The classical concept of recrystallization envisaged the following changes taking place simultaneously as a result of the formation of new grains:

(*a*) A release of stored energy of cold work in proportion to the volume of cold-worked matrix replaced by new, unstrained grains.

(*b*) Softening, as work-hardened material is replaced by soft, annealed grains.

(*c*) Local reorientation, as ordinary high-angle grain boundaries migrate and a deformed grain of a certain orientation is replaced by an annealed grain of a different orientation.

It has been established, however, that these processes do not occur simultaneously.[19, 20] In view of the variable and sometimes negligible amount of strain energy and residual work-hardening left in the matrix, when it is being absorbed by new grains, primary recrystallization cannot be separated from secondary recrystallization or grain growth.

A recent investigation by Borelius [21] has shown that energy can be released before recrystallization, and Suzuki [22] found that the energy release was complete when softening, as measured by indentation hardness, began. Other workers found that the energy release, softening, and recrystallization coincided. From available evidence, it may be concluded that energy release and recrystallization may not be coincident.

The process of recrystallization may be regarded as a phase change of the form:

$$\text{Deformed state} \rightleftharpoons \text{Recrystallized state}$$

Unlike a true phase change, the reaction does not occur at any fixed temperature. The deformed state is metastable at all temperatures, although the rate of transformation may be so slow as to be virtually zero at lower temperatures.

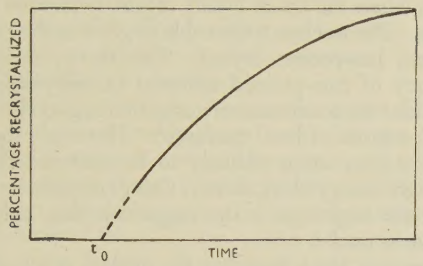

FIG. 4.—Variation of Amount of Recrystallized Material with Time at Constant Temperature.

Fig. 4 illustrates the general kinetics of the recrystallization process, and microscopic examination of grain structure has shown that isothermal recrystallization can be described in terms of a nucleation frequency and linear growth rate. The kinetics has been considered theoretically by several workers.[23, 24] The best-known result is the Avrami expression:

$$x = 1 - e^{-Bt^K}$$

where x is the fraction recrystallized, t is the time, and B is a constant dependent on f, N, and G, which are, respectively, a shape factor, the nucleation frequency, and the linear growth rate. K is also a constant and can assume the following values: 3–4 for three-dimensional grain growth in massive material, 2–3 for two-dimensional growth as platelets in thin strip, and 1–2 for uni-dimensional growth as fibres in wire.

Krupkowski and Balicki[25] have proposed a first-order equation of the form:

$$x = 1 - e^{-Kt}$$

where, as before, x is the fraction recrystallized, t the time, and K the velocity constant. They visualize a homogeneous process of the same form as found for order–disorder reactions in solid solutions. This theory is incompatible with the N–G concept, unless these decrease arbitrarily with time; it may also be rejected, since the initial rate of change is not the greatest, as predicted by the theory.

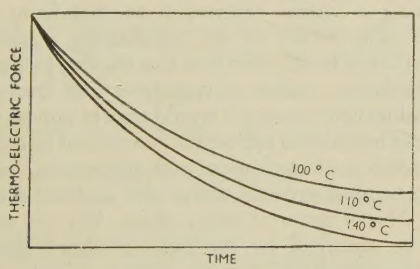

FIG. 5.—Variation of Thermo-e.m.f. with Annealing Time at 3 Different Temperatures.

Using Brandsma's[26] data, Brindley[27] finds that the process is single-stage, as suggested by Krupkowski and Balicki. Brindley followed changes in electromotive force (see Fig. 5), but was not prepared to assert that these were due to recrystallization, as they have been shown to result from a disappearance of work-hardening. It was observed that recrystallization, determined by X-rays, did not vary linearly with the electromotive force. Richards[28] has suggested that the changes are due to recovery. This is feasible, since the curves are of the type obtained for recovery. Brindley and Krupkowski and Balicki are of the opinion that the activation energy of the process is constant, whereas Cottrell[29] favours a decreasing one for recovery.

Cook and Richards[30] have suggested a two-stage mechanism of recovery and recrystallization. The recovery stage is regarded as the removal of lattice distortion over a restricted range. When recovery is well advanced, a second atomic rearrangement produces a recrystallized nucleus. The two processes involve similar atomic rearrangements requiring the initial activation of a small group of atoms, but differ in the range over which they occur. The rate of the second process is governed by the extent to which the first has progressed. Cook and Richards' treatment leads to the expression:

$$\log t = \tfrac{1}{2} \log \log \frac{1}{1 - x} + \text{constant}$$

The recovery period, in this theory, does not produce softening and may even cause hardening, according to these authors.[30] This doubtful postulate is required to account for the appearance of some recrystallized material without any appreciable softening. The mathematical treatment is based on the simplifying assumption that the amount recovered is small, and it is this approximation which leads to the t^2 term. Thorley[31] has shown that this approximation cannot be valid and has given a complete treatment which leads to an expression:

$$\log \frac{1}{1 - x} = A(t - t_0)$$

where A and t_0 are constants.

This treatment allows the activation energy of the two stages to be calculated, and Thorley[31] has done so, using Cook and Richards' results.[30] Although the values of the two activation energies are similar for both stages, Thorley suggests that while recovery and recrystallization may be the same energetically, and in this sense the whole anneal could be called a single-stage process, physically there would still be two manifestations of this single activation energy, namely removal of strain and growth of crystallites from freshly formed nuclei.

Varley,[32] working with aluminium, found a continuous curve and observed some softening without recrystallization. It was suggested that these results are consistent with the theory that recovery and recrystallization take place by diffusion of dislocations in the metal lattice and the mutual annihilation of those of opposite sign. No significant change in activation energy was reported over the annealing period, and it seems likely that, if a new crystal is formed, the additional energy involved in forming it is too small to be observed.

Richards[33] has criticized this conclusion, that a constant activation energy precludes two stages, on the grounds that different physical processes may have the same activation energy, but this seems an artificial distinction to draw, when both recovery and recrystallization or the movement of interfaces can be represented by a dislocation-diffusion mechanism. In this sense the process is a single-stage one.

Several attempts have been made to verify the nucleation-and-growth concept of the process. Karnop and Sachs[34] have shown the growth rate of the grains to be linear. The impingement theory, allowing for the interference of growing grains, has been tested by Mehl and his co-workers,[35] who

found that G was independent of time and N an exponential function of time. Stanley [35] worked with iron and Phillips and Phillips,[36] using heavily deformed copper, have confirmed these findings. Mehl obtained similar results with aluminium.[37]

Cook and Richards' expression is of the same form as Avrami's, and Burke and Turnbull [38] think it possible, even if the recrystallization observed by Cook and Richards was three-dimensional, that the growth rate decreases with time in such a way that $K = 2$. Thus, their results would be consistent with the nucleation-and-growth picture, and it would be unnecessary to invoke a two-stage mechanism. However, the growth rate is independent of time, and it would seem, therefore, that the nucleation-and-growth concept is inconsistent with Cook and Richards' results, unless the recrystallization was in fact two-dimensional.

Furthermore, Thorley [31] and Phillips and Phillips [36] favour an equation which, for long times, assumes the form:

$$x = 1 - e^{A(t - 1/B)}$$

By comparison with Avrami's expression, $K = 1$, which would imply unidirectional growth, a most unlikely event in strip material.

Although the nucleation-and-growth concept appears to be satisfactory for lightly deformed aluminium, it does not offer so satisfactory an explanation for the process in heavily deformed copper. Brindley,[27] working at low reductions of copper, favoured an equation involving the first power of time, although the observations may have been on recovery, and the nucleation-and-growth concept seems to be inapplicable in this case too.

Nucleation

One of the greatest obstacles to a fuller understanding of recrystallization lies in describing the exact mechanism of nucleation. The growth of grains is well known, and a fairly advanced quantitative theory exists, but how these grains are initiated is not so well understood. Several theories of recrystallization nuclei exist, and these will be discussed. Some of the principal facts such theories must account for are:

(1) Nucleation appears more prevalent in areas of higher distortion.

(2) With few exceptions, the activation energy of nucleation does not differ perceptibly from that of grain-boundary migration.

(3) The nucleation frequency increases sharply with strain.

There are two general theories of recrystallization nuclei: (*a*) the conventional nucleation theory, which supposes that nuclei of a stable critical size form most readily in highly strained parts of the matrix, by a sequence of thermal fluctuations, and, having a lower energy, then consume the matrix; and (*b*) the block theory, which regards the nuclei as preformed blocks of material which grow into the matrix, as recrystallized grains.

Conventional theory, according to Burke and Turnbull,[38] requires a local elastic strain of 0·2 to account for observed nucleation frequencies. This, admittedly, is a local strain, but since the amount of strain energy left at the onset of recrystallization is probably small, the theory may not be generally applicable.

The block hypotheses fall into two classes, based upon the energy of the block. The high-energy blocks, which are associated with areas of greatest distortion, become nuclei by a unit process which progressively relieves the whole strain in the block. The nucleation period is associated with the growth of the block to microscopic dimensions, or it may be a period during which recovery takes place in the block. One theory of this type is due to Cahn.[39] Local curvatures arise in deformed metal lattices, and these serve as potential sites for nucleation, where the lattice locally regains perfection by polygonization. The nucleus is then able to grow at the expense of the adjacent, less-perfect crystal. This theory accounts for the tendency of fine-grained material to recrystallize more rapidly and for the occurrence of nucleation at grain boundaries, these being regions of local curvature. However, the process of polygonization seems unlikely to be activated thermally, and, as a high-energy block theory, Cahn's hypothesis becomes invalid; more important is the suggestion that polygonized blocks serve as nuclei.

In low-energy block theories, the nucleus is a perfect area and the incubation period is the time of initial slower growth rate. The rate may be expected to increase with increasing size as the loss of energy for a small transformed volume becomes considerably larger than the gain due to increased interfacial energy. Cottrell [40] has proposed an additional reason for the slow initial growth rate, in that the nucleus is only slightly disoriented from the surrounding matrix, but, as it grows, the boundaries reach regions where the orientation is quite different. Accepting this description of the incubation period, Cahn's suggestion of polygonized blocks acting as nuclei becomes acceptable, since it is based rather on their growth than on their formation.

The polygonized blocks are not initially completely strain-free. With short annealing times, they become more perfect and more easily distinguished. Perryman [15] has observed softening in super-purity aluminium during this period and concludes that it is probably due to the diffusion of dislocations, some cancelling out and others going to the sub-boundaries, thereby increasing the disorientation and accounting for the greater ease of observation. Bearing in mind subgrain behaviour and the occurrence of recrystallization in strain-free material, it must be concluded that Cahn's theory, together with Cottrell's explanation of the slow growth rate, affords the best available explanation of the formation of nuclei. Support is offered by the observation of fine subgrains of similar orientation to the main grain in large grains of aluminium produced by the strain-and-anneal technique. It is not to be expected that the recrystallized grain can absorb those subgrains having a similar orientation to that from which it originated. The virtue of this mechanism is that it presupposes no lattice strain other than that implied by the presence of sub-boundaries; rather, it requires a low level of strain, since the dislocations causing it would tend to oppose boundary migration, as boundaries oppose the movement of dislocations.

The association of nucleation with high-stress zones is due to the more rapid growth where the nucleus is more disoriented. Nuclei grow at other places, but not so rapidly, and may be absorbed by the faster-growing grains. The nucleation frequency will increase with strain, since there will be more distorted regions and local disorientations.

Conclusions

It may be stated that the softening of cold-worked metals which may occur by recovery, especially when the deformation is simple, and by recrystallization, is variable, and that copper may be the only metal softening completely by recrystallization.

The general kinetics of recovery has been established and is

readily interpreted in terms of mutual annihilation of dislocations. Microstructural observations show that recrystallization can be described by the nucleation-and-growth concept, but this is not consistent with other approaches based on the change of some structure-sensitive property. The most attractive of the latter is the two-stage recovery–recrystallization theory. Unfortunately, this theory has been applied only to copper, and no account is taken in it of the sometimes considerable softening that occurs by recovery.

Metals have been shown to consist after deformation of fine subgrains, and it is probably the growth of these which gives rise to polygonization and recrystallization *in situ*. Recrystallization is thought to occur by the growth of a few subgrains, the nucleation period being occupied by an initial slow growth rate or recovery of the subgrains. Such a theory is capable of explaining most known observations regarding nucleation sites and textures. Whatever be the exact extent of recrystallization in relieving strain, some prior degree of recovery appears to be essential to the recrystallization process.

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